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Al-Pd-Mn icosahedral quasicrystal: deformation mechanisms in the brittle domain.

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The extreme brittleness of Al-Pd-Mn quasi-crystalline alloys over a wide range of temperatures drastically restricts the investigation of their plastic deformation mechanisms over a small high-temperature regime. Recently, plastic deformation of Al-Pd-Mn quasicrystal has been achieved in the brittle domain (20°C ≤ T ≤ 690°C) using specific deformation devices, which combined an uniaxial compression deformation or a shear deformation with an hydrostatic pressure confinement (0.35 GPa – 5 GPa). Results obtained with these experimental techniques, which provide various deformation conditions giving rise to a range of Al-Pd-Mn plastic features in the brittle domain, are discussed. On this basis, we propose that low and intermediate temperature plastic properties of Al-Pd-Mn are controlled by non-planar dislocation core extensions that are specific to the non-periodic structure.

Keywords: quasicrystal, dislocation, plasticity, core spreading, confining pressure.

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§1.Introduction

The discovery of an ordered metallic phase with a non-periodic structure [1] induced a great interest in the scientific community. Quasiperiodic structures have been described using mathematical method [2,3] and appropriate indexation system has been built up for icosahedral quasicrystals [4]. Theoretical predictions of their structural defects and corresponding extinction conditions in transmission electron microscopy (TEM) were proposed [5] and verified by experimental observations [6].

The mechanical properties of quasicrystalline alloys are characterized by an extreme brittleness at low and intermediate temperatures, corresponding to a high brittle-to-ductile transition (BDT) temperature usually close to 70% of their melting temperature. This behaviour strongly suggests that diffusion plays an essential role in quasicrystal plasticity. Activation volumes, which were measured at high temperatures using different methods [7-11], have values too high to correspond to a deformation mechanism that would only imply diffusion. The stress-strain curves obtained at high temperatures for conventional deformation tests exhibit a yield point similar to the one observed in semi-conductors. For Si and Ge, the origin of the yield point has been ascribed to a lack of mobile dislocations [12], while for Al-Cu-Fe poly-quasicrystals it has been attributed to a lack of dislocation mobility [13]. Cottrell-Stokes type experiments [14] performed on Al-Cu-Fe [15] and Al-Pd-Mn [16,17] quasicrystals show that the flow stress is almost fully reversible with deformation temperatures, indicating that the variation of the applied stress with temperature essentially reflects the variation of the effective stress.

Dislocation activity has been observed in Al-Pd-Mn quasicrystalline thin foils deformed at high temperature during *in situ* TEM experiments [18], suggesting that plastic deformation may be ascribed to dislocation motion. However, the microscopic mechanism, by which dislocations move, has not been properly identified and the rate controlling mechanism is still a subject of debate. Previous observations of high-temperature deformed Al-Pd-Mn alloys concluded that dislocation motion essentially occurred by glide [19], whereas more recent studies of as-grown [20, 21] and deformed specimens [22] both give evidences of predominant climb processes.

These contradictory results deserve further study for identifying the microscopic deformation mechanism that controls quasicrystal plasticity. In particular, low temperature deformation, *i.e.* when diffusion is unlikely to occur, may significantly contribute to elucidate if pure dislocation glide can be initiated. Al-Pd-Mn specimens were therefore deformed at low and intermediate temperatures, using experimental techniques specifically designed for brittle materials, and the related microstructures were examined by TEM. Special attention was paid to obtain from the *post mortem* observations possible dislocation kinetics at the origin of the observed dislocation configurations and to estimate the respective contribution of dislocation glide and dislocation climb to plastic deformation. The peculiar plastic behaviour of Al-Pd-Mn is explained and discussed in terms of a model based on non-planar core dislocation configurations.

§.II. Experimental conditions

Poly-quasicrystals and mono-quasicrystals of nominal composition $Al_{66.3}Pd_{21.9}Mn_{11.8}$ and $Al_{70.5}Pd_{21.0}Mn_{8.5}$, respectively, were deformed under various deformation conditions. Poly-quasicrystalline structure, which promotes homogeneous plastic deformation for all deformation systems, was used in a first attempt to determine the main characteristics of Al-Pd-Mn plasticity. The mono-quasicrystalline specimens, which allow the initiation of plastic flow along particular directions and planes, were used to verify special features suggested by the study of poly-quasicrystalline samples. For all specimens, plastic deformation was applied via compression or shearing tests under confining pressure produced either by a solid (multi-anvils, at 5 GPa) or gaseous medium (Paterson press, at 300-400 MPa) to limit crack nucleation and propagation in the brittle specimens. One must notice that stress-strain curves are only available for the latter type of experiment. Performances and limitations of the various deformation devices have been reported in detail in reference [23]. Compression and shearing tests were performed at constant strain-rate, $\dot{\varepsilon} = 1.5 \times 10^{-5}\ s^{-1}$, over the temperature range 20°C - 690°C. Description of compression and shearing devices together with specimen preparation and detailed deformation conditions can be found elsewhere [24, 25]; they are summarised in table 1.

Thin foils for TEM were first cut from the deformed specimens using diamond saw and subsequently thinned by ion milling at liquid nitrogen temperature. The microstructures were observed in 'two-beam' conditions by TEM with a JEOL 200 CX transmission electron microscope operating at 200 kV.

§.III. Experimental results

III.1. Macroscopic behaviour

The stress-strain curves obtained from both the compression and the shearing tests under gaseous confining pressure are similar to those obtained from conventional deformation tests performed at high temperature and atmospheric pressure. After a few percent of plastic strain, the applied stress reaches a maximum value, referred to as σ_{UYS}, followed by a softening stage of continuous stress decrease. Examples of stress-strain curves obtained for compression tests are presented in figure 1a, similar curves for the shearing tests are available elsewhere [26].

The temperature dependence of σ_{UYS} is reported in figure 1b for the two types of test, together with literature σ_{UYS} data obtained from conventional compression experiments performed at high temperatures [27]. One can remark, except at 690 °C where experimental difficulties were encountered (see figure 1a; the 690 °C compression test was disturbed at 4% total strain by a temporary pressure leak leading to an artefact in the load cell signal, as seen on the stress-strain curve), that the σ_{UYS} -T dependence nicely follows a linear fit over a large temperature range, which for the imposed strain-rate includes temperatures that belong to the brittle and ductile domains. One should also note that a linear fit was also obtained only based on high temperature results [28].

III.2. Microstructures

III.2.1. Low temperature deformation (T<480 °C)

Detailed analyses of the low temperature deformation microstructures have already been given elsewhere [24, 29, 30]. They can be summarized as follows.

Microstructural observations of specimens deformed at low temperature show that plastic deformation is, as a rule, confined to very localized regions. Numerous straight bands containing high dislocation densities characterize the microstructures. Careful examination of the dislocations located at the tip of the bands indicates that they predominantly have a screw character, their curvatures suggesting that they have moved by glide in 5-fold planes. Planar defects bounded by partial dislocations are also observed at all temperatures, with a density that clearly increases with increasing temperature. Determination of fault habit planes together with related $b_{//}$ Burgers vector directions indicates that these dislocations have moved either by climb or by a process that involves a large climb component. In an attempt to check the stability of these faults, *in situ* heating experiments of deformed specimens have been performed in TEM. The planar faults were stable even after a heating of several tens of minutes at 700 °C.

III.2.2. Compression tests at intermediate temperatures (480 °C ≤ T ≤ 690 °C)

The dislocation microstructures of samples deformed at 480°C and 690°C exhibit some common features, which are:

- dislocations are quite homogeneously distributed in the thin foils,
- tri-dimensional networks of reacting dislocations are frequently observed,
- dislocations are mainly located in 5-fold planes and show a segmented aspect, the dislocation segments being preferentially oriented along two-fold directions,
- dislocation dipoles are frequently observed. A complete description of such dislocation dipoles has been given in [25].
- isolated planar defects bounded by partials dislocations are still observed, but they are quite rare and do not constitute anymore a dominant feature of the deformation microstructures.

Figure 2a shows a typical example of segmented dislocations observed in deformed specimens at 690°C. The dislocation density is rather low, suggesting that their motion mainly results from the imposed stress and was not strongly influenced by dislocation interactions. Different imaging conditions have been used for determining their main characteristics. The rectilinear portions of each dislocation are aligned along 2-fold directions. The segment directions, labelled from 1 to 4 in figure 2a, lie in a common five-fold plane, which is perpendicular to the compression axis. All dislocations in figure 2a have the same extinction conditions. The physical component of their Burgers vectors is found to be parallel to a 2-fold axis, inclined at 32° from the habit plane normal. The dislocation segments have either a mixed character (segments 1 to 3) or a pure edge character (segment 4). It must be emphasised that, in contrast to low deformation temperatures, no dislocation with a pure screw character has been observed. Figure 2b is a schematic of one of the observed dislocation configurations in figure 2a. The glide plane, defined by the line direction and the physical component $b_{//}$ of the Burgers vector, is shown for each segment. It can be seen that all segments, which belong to the same five-fold plane, have different glide planes, forming a glide cylinder. Note that this configuration leads to different Schmid factors for each dislocation segments when glide is considered. Assuming that dislocation mobility for glide is dependent on the plane of motion, this geometrical configuration strongly suggests that the expansion of the dislocation loop occurred in the common five-fold plane, which contains all the dislocation segments. Since $b_{//}$ makes an angle of 32° to the normal of this plane, these dislocations have moved by a mechanism that includes a large climb component.

II.2.3. Shear experiments at intermediate temperatures (480 °C ≤ T ≤ 690 °C)

The deformation microstructures of sheared mono-quasicrystalline specimens depend strongly on the deformation temperature. The dislocation microstructure of specimens sheared at 480 °C is heterogeneously distributed within the thin foils, as confirmed by the observation of specimen surfaces performed by scanning electron microscopy [26], where some areas contain high densities of cracks. In these regions, no clear evidence of dislocation activity is detected. Other zones, which certainly correspond to larger shear amounts, are fragmented in numerous disoriented grains of few hundreds nanometres in size (see Figure 3a). The corresponding selected area diffraction patterns exhibit isolated spots distributed over discontinuous rings (see Figure 3b), which confirm the small grain poly-quasicrystalline structure. It must be emphasized that periodically aligned spots are observed (see for instance white arrows in figure 3b). This bears witness to the presence of one or several grains with a crystalline structure. Since the specimens were initially of highly perfect quasicrystalline phase [28], we conclude that phase transformation has been induced by the deformation test. Pressure-induced phase transformation has already been reported for bulk Al-Pd-Mn [31] and, more recently, for nano-indented Al-Pd-Mn single quasicrystals [32].

The microstructure of specimen sheared at 690 °C presents similar characteristics to the one obtained by compression at the same temperature. The dislocation microstructure is homogeneously distributed in all the investigated thin foils. Stereographic analyses indicate that the dislocations are contained in 5-fold planes, for which the shear stress is maximum. Like for the dislocations observed in samples deformed by compression test, a careful examination of the dislocation configurations strongly suggest that their motion occurred by a climb process. Therefore, in spite of the very favourable deformation conditions for promoting dislocation glide (shear plane parallel to a 5-fold plane), no evidence of dislocation glide has been identified in the specimens sheared at 480°C and 690°C.

§.IV. Discussion

IV.1. Dislocation features

With the exception of the specific case at room temperature, where screw dislocation pile-ups are probably moving by glide, all other considered dislocation configurations suggest that their formation results from dislocation movements that imply a predominant climb component. In addition, experiments specifically designed for promoting dislocation glide, *i.e.* shear experiment under high confining hydrostatic pressure, failed to provide evidence of dislocation configurations that would result from glide processes. Therefore, the present results confirm that dislocation glide is extremely difficult to initiate in this quasicrystalline phase, at least for temperatures that belong to the brittle domain. Nevertheless, on the basis of our observations only, we cannot definitively conclude that dislocation glide does not take place in the icosahedral phase, as was claimed for instance by others authors [33]. Dislocation motion may strongly depend on the plane of motion, which would require a more extensive study where not only 5-fold planes are preferentially favoured. In addition, the amount of plastic strain may also have some importance, as explained below, and careful investigations at various strain levels must be carried out.

The strong propensity of dislocations to move by non-conservative mechanisms certainly constitutes the dominant characteristic of Al-Pd-Mn plasticity at low and intermediate temperatures. In particular, diffusion processes are unexpected to occur at such temperatures. Dislocation climb is usually expected to play an important role, generally ascribed to dislocation microstructure recovery, when deformation temperature reaches values higher than nearly one half of the melting temperature. In the present case, dislocation climb does not contribute to dynamic recovery only but also constitutes the main process by which plastic deformation is produced. This type of climb-controlled plasticity is, in some aspects, similar to the one of the hexagonal structure, such as Be and Mg, deformed under specific loading conditions [34].

The dislocation characteristics also exhibit some unexpected changes when the deformation temperature is raised from low, *i.e.* typically room temperature, to intermediate temperatures, *i.e.* essentially below the brittle-to-ductile transition temperature. At low temperatures, some dislocations display characteristic typical fault contrasts indicating that they participated to the plastic deformation. The fringe contrasts associated with these planar defects do not vanish during *in situ* TEM heating of deformed specimens, as observed for pure phasonic faults [35]. These planar defects have a complex nature, which includes a topologic component in the quasiperiodic stacking sequence together with a phasonic component. Consequently, these dislocations are partial dislocations in the 6-dimensionnal (6D) space, *i.e.* their 6D Burgers vectors are not translational vectors of the 6D periodic hyperlattice, E_{6D} . At intermediate temperatures, dislocations bounding planar defects are very rare. As a rule, the dislocations do not present fringe contrasts in their vicinity anymore, indicating that they now have 6D Burgers vectors which are perfect translation vectors in E_{6D}. This also indicates that the inherent phasonic faults dragged by moving dislocations have been reordered during the deformation experiments, leading to the recovery of the perfect icosahedral structure.

Another significant difference with increasing deformation temperature concerns the shape of the dislocation lines, which evolves from curled aspects, with no preferential orientation, at low temperatures to segmented configurations, with dislocation segments aligned along particular quasi-crystallographic directions, at intermediate temperatures. Three main hypotheses are usually proposed to explain the latter dislocation shapes, which are:

• *elastic constant anisotropy* that leads to dislocation line alignments along energetically favourable directions. However, the high isotropy of the icosahedral structure does not support *a priori* such an assumption.

• *Peierls friction force*, the origin of which arises from the influence of the lattice periodicity on a moving dislocation. In order to be efficient, this friction force necessitates strong directional atomic bondings and leads to dislocation alignment along specific directions that usually correspond to close-packed atomic directions. In Al-Pd-Mn quasicrytals, such preferential directions are two-fold quasicrystallographic axes. The segmented aspect is more and more pronounced with increasing temperature and must be interpreted in connection with the viscous dislocation motion observed during *in situ* TEM observations [18]. For a glide process, Peierls forces are believed to be overcome by the thermally activated nucleation of kink pairs and the subsequent sidewise motion of the kinks [36], akin to what has been proposed for crystalline solids [37]. However in the present study, whatever the deformation temperature is, dislocation motion always implies a climb component, ruling out this classical description. The concept of Peierls forces [38] must be extended in the frame of a non-conservative dislocation movement that replaces kink by jog, although the concept of Peierls forces is not appropriate in the case of climb [P. Beauchamp, private communication]. Since Al-Pd-Mn plasticity is climb controlled, dislocation multiplication occurs through Bardeen-Hering type source, which precludes pre-existing jogs and, therefore, requires jog pair nucleation. The similarity between kink pair and jog pair nucleation and migration has been used to explain segmented configuration [39]. In this description, a rectilinear dislocation aspect is obtained when the time for jog pair nucleation is larger than the time for jog migration over distances equal to the mean inter-jog spacing, whereas the reverse situation leads to curled dislocations. The evolution of dislocation shapes from curled to segmented would imply that jog-pair nucleation is, at low temperature, a process easier than jog migration, with a reversal in the respective ease with which the two processes occur with increasing temperatures. Based on the simple assumptions that both mechanisms require diffusion and that only jog pair nucleation increases dislocation energy, one should expect that, whatever the temperature is, jog pair nucleation is more difficult than jog migration. In addition, dislocation cores can be seen as lines of easy diffusion, similar to dislocation core in crystalline structure, and enhanced jog mobility. This should be true particularly at low temperature, where bulk diffusion can be considered as negligible. Consequently, while a competition between jog pair nucleation and jog migration cannot be entirely refuted for explaining the dislocation shape and its evolution with temperature, this interpretation appears rather unlikely.

- *dislocation core extensions* out of the plane of motion, in particular when dislocation core extends in more than one crystallographic plane. In this case, the dislocation tends to align along the common intersection line of the dissociation planes such as, for instance, screw dislocations in bcc crystals. Then, the segmented aspect results from energetically favourable dislocation configurations, which are similar, by some aspect, to Peierls valleys. However, dislocation dissociation in several planes yields sessile dislocation configurations, which are usually considered as impeding dislocation motion. In this description, we expect that an increase in core extension with increasing temperature would result in a decrease in dislocation mobility, which contrasts with the increasing dislocation activity observed when temperature is raised.

Nevertheless, as shown below, the temperature evolution of dislocation core gives a reasonable explanation for the evolution of dislocation shape with temperature.

IV.2. Interpretation of dislocation motion in icosahedral quasicrystals

IV.2.1. Displacement field associated with dislocations in quasicrystals

Dislocations can be introduced in quasicrystals by a generalization of the Volterra process in a 6-dimensional space, E_{6D}, where the full translational symmetry is obeyed. They can be described as singularities in the periodic hyperlattice, characterized by a 6D Burgers vector $\vec{B}$ [5]. The 6D Burgers vector $\vec{B}$ corresponds to the integration around the displacement field of the 6D hyperlattice, which can be separated into an elastic distortion ($b_{//}$) and a phason distortion ($b_{\perp}$). The complete displacement field of a dislocation may then be expressed as a function of $\vec{b}_{//}$ and $\vec{b}_{\perp}$, which describe the quasiperiodic lattice elastic distortion (phononic displacement field) and the rearrangement of quasiperiodic lattice nodes (phasonic displacement field), respectively. Since the quasicrystalline structure resulting from the '*cut and project*' method does not depend on the choice of the origin in $E_{\perp}$, a shift of the projection space $E_{//}$ parallel to $E_{\perp}$ has no effect on the overall quasicrystalline structure. However, a local displacement field in $E_{\perp}$ leads to local tiling mismatch in $E_{//}$, that may have different nature (chemical, geometrical or both) depending on the position in $E_{//}$ where the displacement is applied. This displacement depends of the amplitude of $\vec{b}_{\perp}$ with respect to the atomic surface extension in $E_{\perp}$, which varies according to its position in $E_{//}$.

Therefore, the total displacement field in $E_{//}$ due to a dislocation line can be written as the sum of a constant component $\vec{u}_{//,\vec{b}_{//}}$, which is proportional to $\vec{b}_{//}$ and a component $\vec{u}_{//,\vec{b}_{\perp}}$, which varies in direction and amplitude along the dislocation line in $E_{//}$.

An analogy of dislocation displacement field in quasicrystals can be made with the scheme of topological linear defect in solids proposed by Somigliana [40, 41]. In this treatment, linear defects are produced by joining the two surfaces of a cut cylinder, where the two surfaces have been independently deformed prior to the translation along the cut surface. The resulting defects are not characterized by a constant displacement vector as in the case of a Volterra's dislocations: the magnitude and direction of the Burgers vector of these dislocations both fluctuate about an average value $<\vec{b}>$, which depends on the local surrounding material structure. The displacement field is then described by the sum of a constant Burgers vector component, $<\vec{b}>$, and a variable component, arising from localised dislocation loops, which accounts for local fluctuations of the total Burgers vector. In the quasicrystalline structure, the local Burgers vector corresponds to the sum of the $\vec{b}_{//}$ component and a variable component, which represents, in the physical space, the phasonic displacements due to the $\vec{b}_{\perp}$ component. Since the atomic displacements due to this variable component have an equal probability to occur in all directions, the comparison of both dislocation descriptions leads to assimilate $<\vec{b}>$ in amorphous solids to the $\vec{b}_{//}$ component in quasicrystalline structures. In amorphous solids, the motion of dislocations with non-constant Burgers vectors implies the nucleation of small prismatic dislocation loops in front and/or in the wake of the moving dislocation [42], in a similar way as phasons are created in quasicrystals. Like phasons in the case of quasicrystals, the dislocation loops do not directly participate to plastic deformation, but allow accomodating the geometric mismatch between the two interfaces in translation. As a consequence, the deformation is confined to narrow deformation bands, which contain the loops, phasons here, produced by the first moving dislocation. This collective dislocation behaviour is usually associated with a yield point on the σ-ε curves obtained for compression tests performed at constant strain-rate. It also leads to very localised plastic deformation events corresponding to small activation volumes and, in a correlative manner, to a high temperature sensitivity of the yield stress. Amorphous and quasicrystalline materials share all these characteristics.

IV.2.2. Dislocation behaviour

The variable component of the displacement field can produce a variety of atomic displacements along the dislocation line, which for energetic reasons may lead to a spreading of the dislocation core. In Al-Pd-Mn, TEM observations suggest that dislocations have non-planar cores that belong to the “climb” class, as classified in [43]. In this case, the dislocation core spreads into several planes, which belong to the zone of the vector parallel to the dislocation line. The Burgers vectors of the fractional dislocations involved in the core spreading possess components that are perpendicular to spreading planes. The existence of climb cores has been reported in several materials [44-49]. Such dissociations usually imply severe restrictions to dislocation movement, in particular when the temperature is decreased to the level at which dislocation climb is inhibited.

Climb core extension is supposed to influence only edge and mixed dislocation motions and cannot be applied to screw dislocation, for which motion occurs by glide only. It must be noticed that the mobility of screw dislocations can be strongly limited by sessile core extensions. For given projection conditions, the volume extensions of dislocation core are function of the dislocation position in E_{6D} and of the Burgers vector ratio ξ, where $\xi = |b_{\perp}|/ |b_{//}|$. This implies that the motion and the mobility of dislocations are both dependent on their ξ values. As a consequence, the densities of dislocations characterized by different ξ parameters should evolve during plastic deformation, leading to an increase in the density of dislocations with the most energetically favourable ξ ratio. For ξ ratio lower than a critical value $\xi_{c,}$ dislocation movements are controlled by core extensions that do not favour either climb or glide mechanisms, whereas for higher ξ values dislocation climb is strongly enhanced, due to high anisotropic core extensions outside the glide plane as depicted in figure 4. This interpretation is in agreement with the observed evolution in dislocation densities with plastic strain characterized by different ξ values [27, 49].

From TEM observations of dislocation microstructures performed on specimens deformed at high temperatures [27, 49], it appears reasonable to assume that $\xi_c = \tau^5$ corresponds to the critical value above which dislocation motion is restricted to climb. At all temperatures, dislocation core extensions outside the glide plane confine dislocation motion in or near the climb plane, in agreement with the observed dislocation configurations [25, 29, 33].

Dislocation core extensions, which are assumed to result from geometrically necessary atomic displacements along the dislocation line, are dependent on the geometrical disorder of the quasicrystalline structure. The existence of a displacement field component $\vec{u}_{//,\vec{b}_{\perp}}$ corresponds to a decrease in the dislocation misfit energy by partial relaxation the local mismatch produced by the constant $\vec{u}_{//,\vec{b}_{//}}$ component of the perfect quasicrystalline structure. Thus, the contribution of $\vec{u}_{//,\vec{b}_{\perp}}$ with temperature may be significantly reduced by the presence of a higher phasonic defect density. In that case, the amplitude of the core extensions out of the glide plane is reduced. Assuming that the variations in elastic and misfit energies of dislocations in quasicrystals are similar to the ones in crystalline structures [38, 50], the decrease of the core extension in the climb plane yields an increase of the total dislocation energy. This results in a small core spreading in the glide plane, leading to some kind of Peierls forces acting on the dislocation line in the climb plane, which are believed to be at the origin of the observed segmented dislocation configurations.

Since vacancy diffusion, which increases exponentially with temperature, contributes to phasonic defect nucleation and propagation (as schematised in figure 5); an increasing disorder is therefore expected at high temperature, leading to higher dislocation segmentation when temperature is raised.

§ V. Conclusion

In this study, experimental techniques using confining pressure were used to characterise the plastic behaviour of Al-Pd-Mn icosahedral quasicrystals in the brittle domain. At room temperature, dislocation glide is evidenced, but for all deformation temperatures below the BTD temperature, dislocation climb is identified. Above room temperature, this mechanism becomes rapidly predominant with increasing temperatures. At temperatures close to the BTD temperature, dislocations exhibit segmented aspects. This feature is also observed for higher deformation temperatures. We propose that low and intermediate temperature dislocation climb and segmented dislocations have a common origin: a non-planar dislocation core extension, which inhibits dislocation glide and, to a lesser degree, dislocation climb. This non-planar core extension, specific to the non-periodic structure, is assumed to evolve with strain and temperature concomitantly with phasonic defect concentration.

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Table caption

Table 1 : Deformation parameters of the various deformation tests (*C*: composition in at. %, *D. T.* : deformation type, *T*: temperature in °C, *C. M.*: confining medium, *p*: hydrostatic confining pressure in GPa, ε : permanent strain after testing in %).

Samples	*C (in at. %)*	*D.T.*	*T (°C)*	*C. M.*	*p (GPa)*	ε *(%)*
Poly-grained	$Al_{66.3}Pd_{21.9}Mn_{11.8}$	compression	20	Solid	5	~ 6
Poly-grained	$Al_{66.3}Pd_{21.9}Mn_{11.8}$	compression	150	solid	5	~ 6
Poly-grained	$Al_{66.3}Pd_{21.9}Mn_{11.8}$	compression	300	solid	5	~ 6
Single-grained	$Al_{70.5}Pd_{21.0}Mn_{8.5}$	compression	480	gazeous	0.36	~ 5
Poly-grained	$Al_{66.3}Pd_{21.9}Mn_{11.8}$	compression	580	gazeous	0.35	~ 6.5
Poly-grained	$Al_{66.3}Pd_{21.9}Mn_{11.8}$	compression	610	gazeous	0.35	~ 6
Single-grained	$Al_{70.5}Pd_{21.0}Mn_{8.5}$	compression	690	gazeous	0.35	~ 6.5
Single-grained	$Al_{70.5}Pd_{21.0}Mn_{8.5}$	shear	480	gazeous	0.36	>20
Single-grained	$Al_{70.5}Pd_{21.0}Mn_{8.5}$	shear	690	gazeous	0.3	>20

TABLE 1

Figure captions

Figure 1. a) Stress-strain curves obtained during compression tests under a gaseous confining pressure of 0.35 GPa. b) Upper yield stress, σ_{UYS}, as a function of temperature: (o) uniaxial compression tests and (x) shear tests obtained under confining pressure,(■) conventional compression tests at atmospheric pressure (from [27]).

Figure 2. a) Example of dislocations observed in a specimen deformed at 690°C. All dislocation segments are aligned along 2-fold directions. b) Schematic of the corresponding dislocation configuration. Dislocation segments are lying in a common 5-fold plane, but belong to different glide planes: dislocation segments (1) and (2) have two distinct 3-fold glide planes, segment (3) has a 5-fold glide plane.

Figure 3. a) Microstructure of a specimen sheared at 480°C. A microstructure with small disoriented grains is observed. b) Related selected area electron diffraction pattern. The arrows indicate periodically aligned spots.

Figure 4. Schematic of expected ξ dependences of dislocation core extensions:

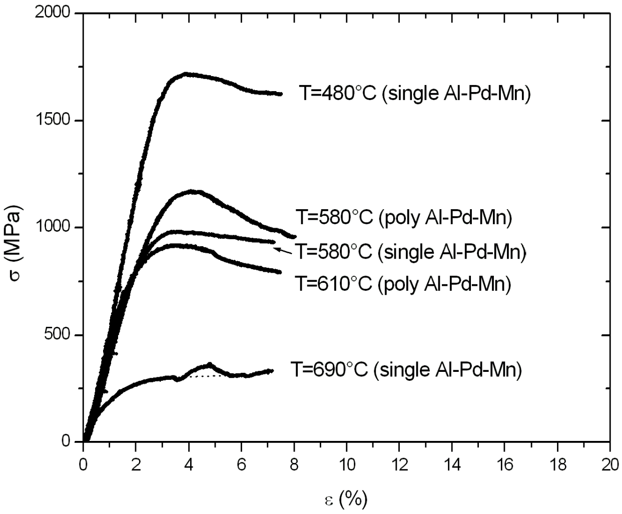
a) Low temperatures.

• for low ξ values, $\vec{u}_{//},\vec{b}_{//}$ and $\vec{u}_{//},\vec{b}_{\perp}$ displacement fields are similar, leading to almost equivalent core extensions in the glide and climb planes.

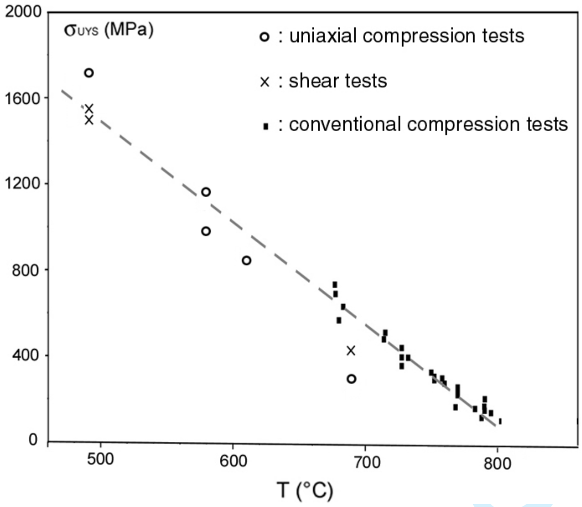
• for high ξ values, extensions out of the glide plane are predominant, promoting dislocation climb.

b) High temperatures. With increasing temperatures, due to the phasonic displacement field, the geometrically necessary atomic displacements are reduced, leading to a decreasing core extension in the climb plane.

Figure 5. Diffusion of a vacancy in a 2D-quasiperiodic structure: (a) the vacancy (open square) occupies a node site of the quasiperiodic lattice. (b) The vacancy and one atom exchange their sites. (c) The vacancy moves away to another position of the quasi-lattice, but the atom sits at an intermediate position (open circle). This site does not correspond to an actual node position of the perfect quasiperiodic lattice and represents a so-called phasonic defect, referred to as ‘basculon’. In that case, vacancy diffusion enhances phasonic defect nucleation, allowing a reduction in the atomic flux associated with vacancy migration.


2000
1500
1000
500
0
0
2
4
6
8
10
12
14
16
18
20
σ (MPa)
ε (%)
T=480°C (single Al-Pd-Mn)
T=580°C (poly Al-Pd-Mn)
T=580°C (single Al-Pd-Mn)
T=610°C (poly Al-Pd-Mn)
T=690°C (single Al-Pd-Mn)

a)


2000
1600
1200
800
400
0
500
600
700
800
σUYS (MPa)
T (°C)
o : uniaxial compression tests
× : shear tests
▪ : conventional compression tests

b)

FIGURE 1

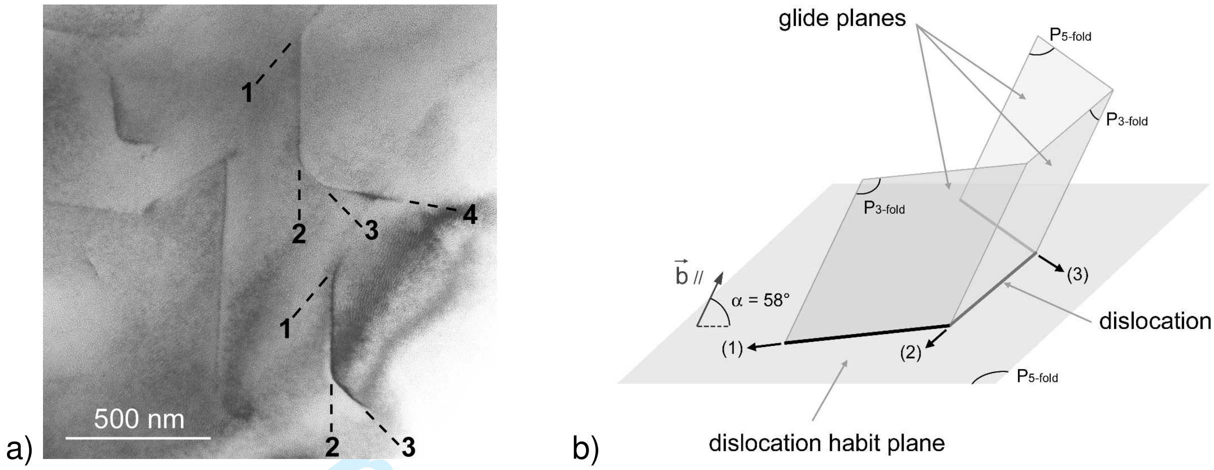

a)
500 nm
1
2
3
4
b)
glide planes
P5-fold
P3-fold
b //
α = 58°
(1)
(2)
(3)
dislocation
dislocation habit plane

FIGURE 2

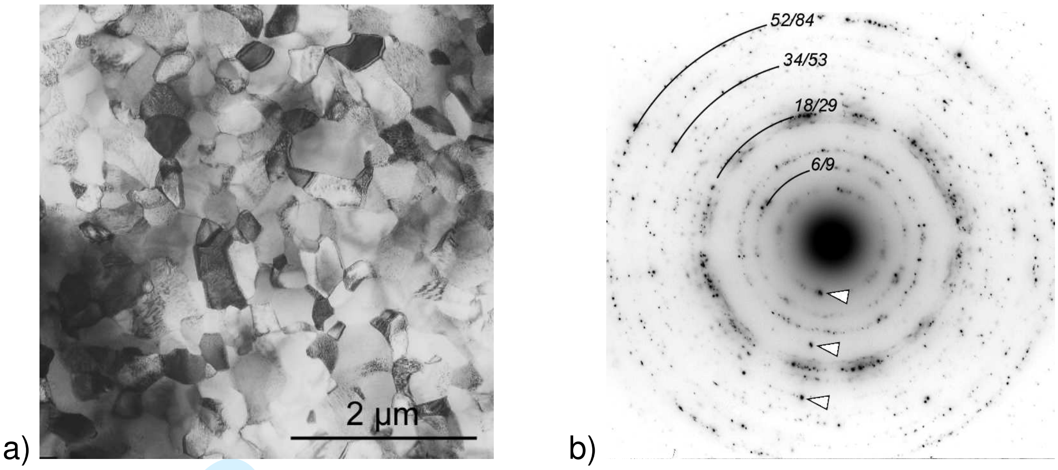


FIGURE 3

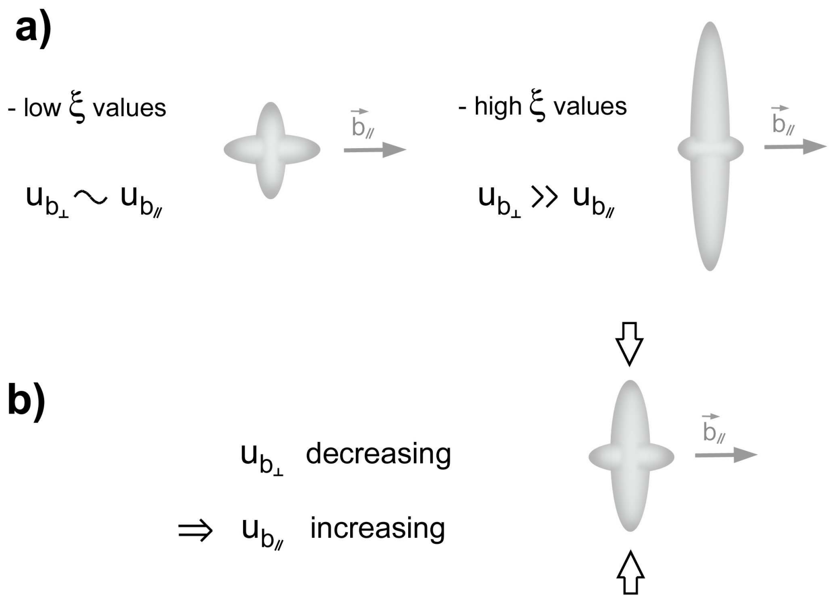

a)
- low ξ values
$u_{b_\perp} \sim u_{b_{/\!/}}$
$\vec{b}_{/\!/}$
- high ξ values
$u_{b_\perp} \gg u_{b_{/\!/}}$
$\vec{b}_{/\!/}$
b)
$u_{b_\perp}$ decreasing
$\Rightarrow u_{b_{/\!/}}$ increasing
$\vec{b}_{/\!/}$

FIGURE 4

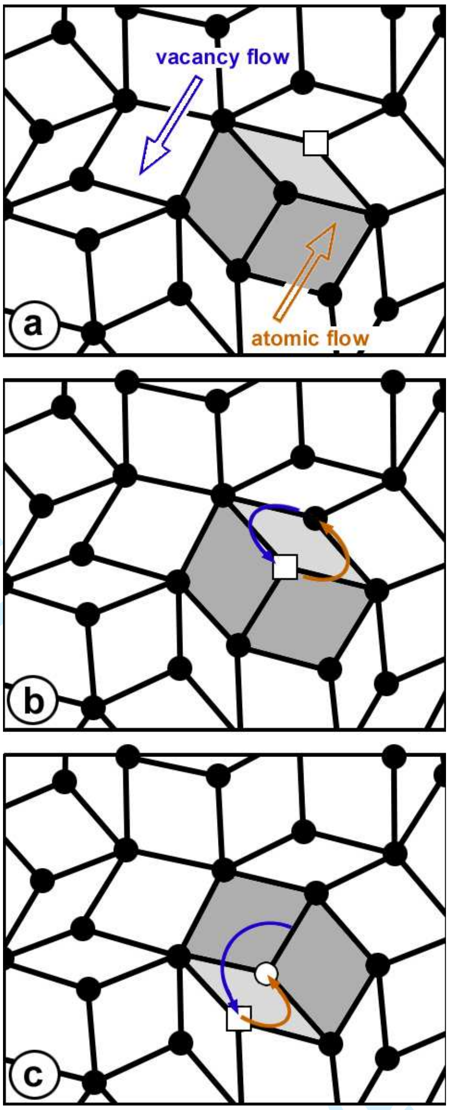
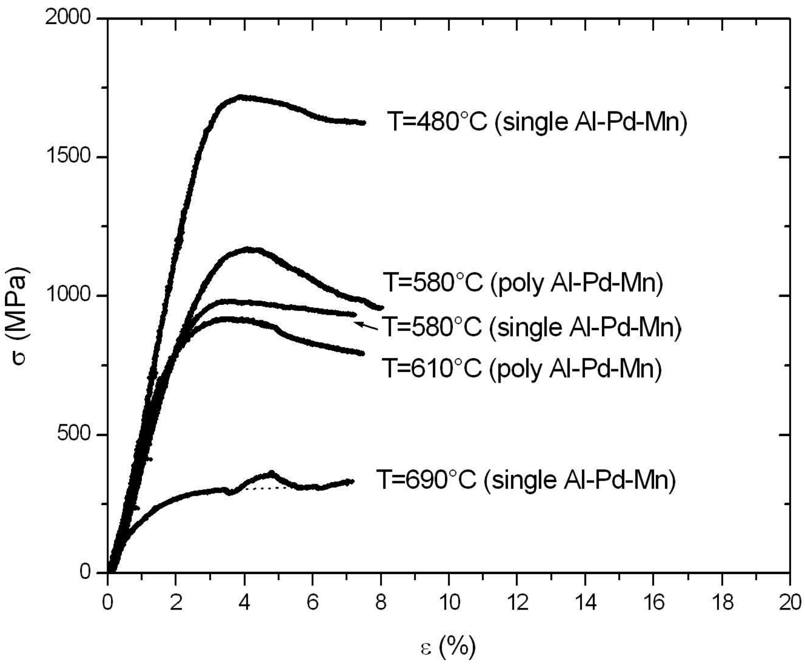
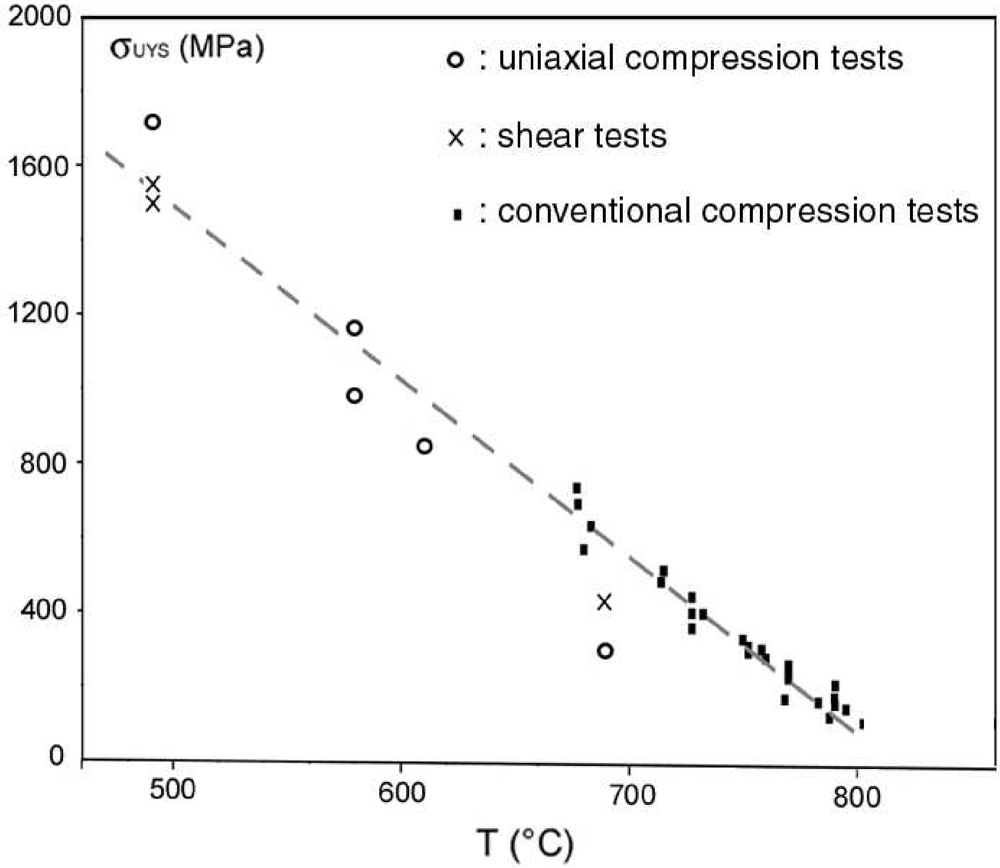


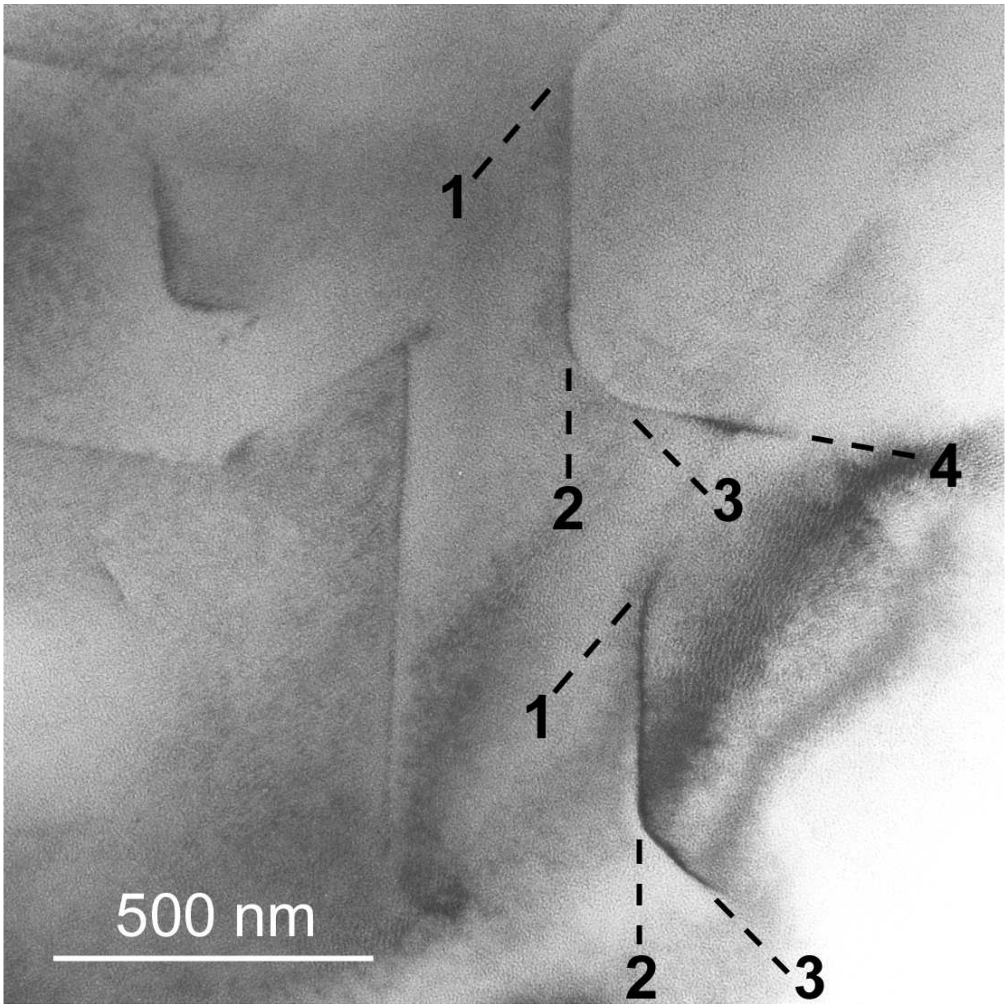
FIGURE 5



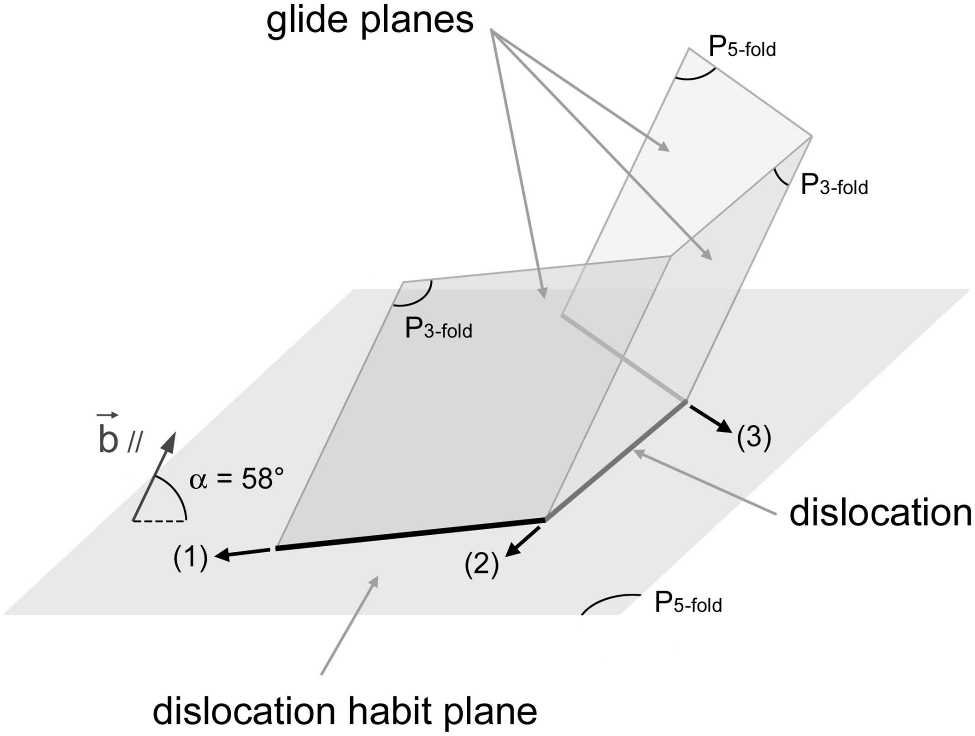
Stress-strain curves obtained during compression tests under a gaseous confining pressure of 0.35 GPa.
279x215mm (150 x 150 DPI)



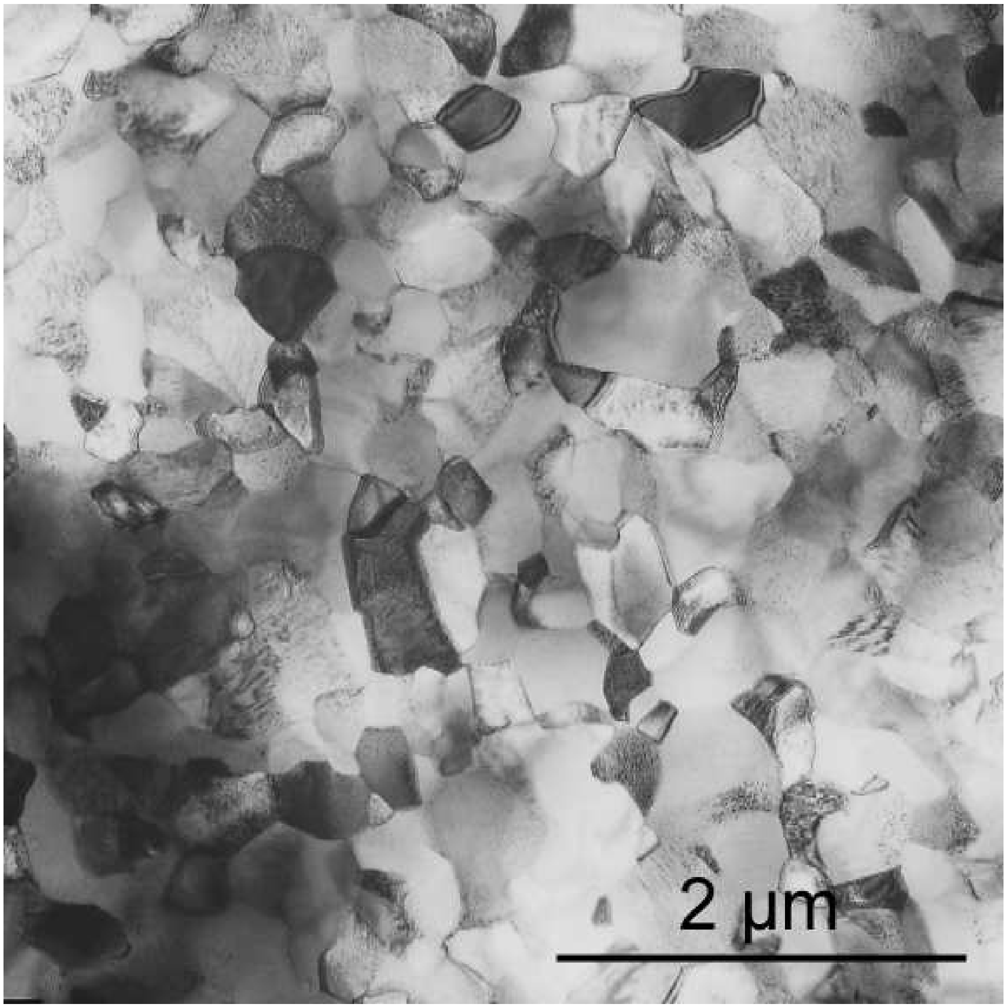
Upper yield stress, σUYS, as a function of temperature: (o) uniaxial compression tests and (x) shear tests obtained under confining pressure,(▪) conventional compression tests at atmospheric pressure (from [27]).
114x99mm (154 x 154 DPI)



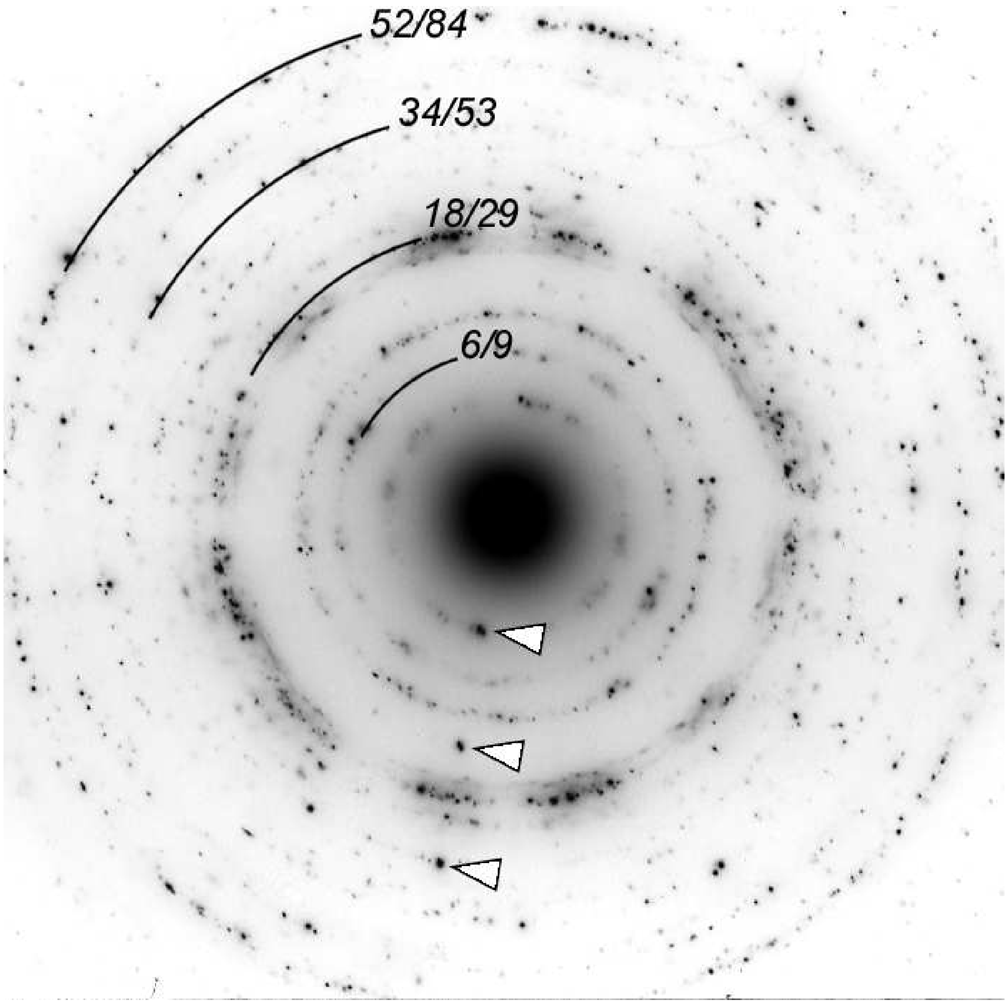
Example of dislocations observed in a specimen deformed at 690°C. All dislocation segments are aligned along 2-fold directions.
46x46mm (600 x 600 DPI)



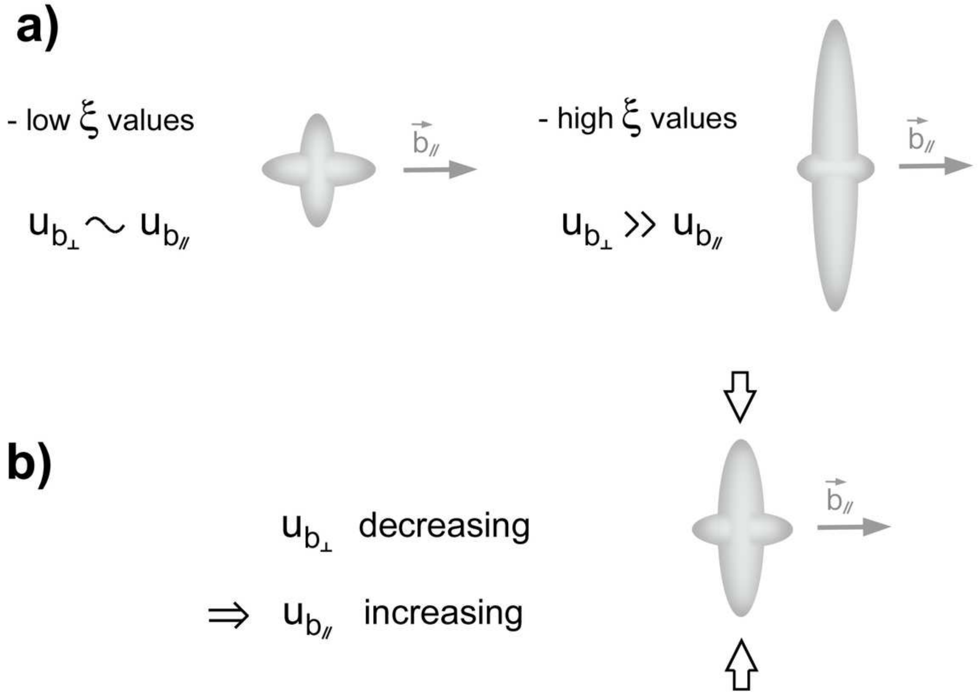
Schematic of the corresponding dislocation configuration. Dislocation segments are lying in a common 5-fold plane, but belong to different glide planes: dislocation segments (1) and (2) have two distinct 3-fold glide planes, segment (3) has a 5-fold glide plane.
144x107mm (381 x 381 DPI)



Microstructure of a specimen sheared at 480°C. A microstructure with small disoriented grains is observed.
48x48mm (300 x 300 DPI)

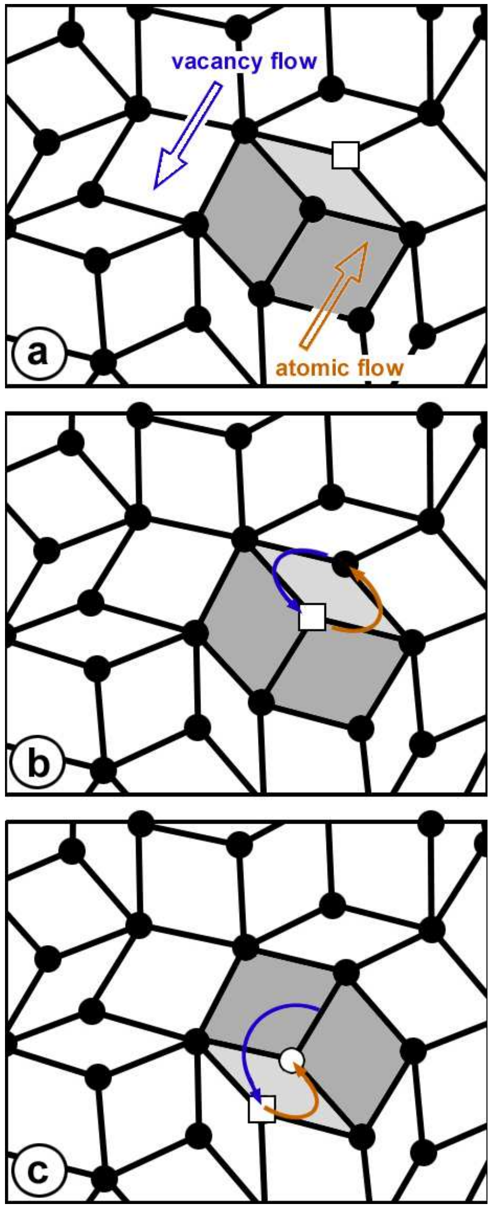


Related selected area electron diffraction pattern. The arrows indicate periodically aligned spots.
59x59mm (300 x 300 DPI)



Schematic of expected ξ dependences of dislocation core extensions:
a) Low temperatures.
• for low ξ values, and displacement fields are similar, leading to almost equivalent core extensions in the glide and climb planes.
• for high ξ values, extensions out of the glide plane are predominant, promoting dislocation climb.
b) High temperatures. With increasing temperatures, due to the phasonic displacement field, the geometrically necessary atomic displacements are reduced, leading to a decreasing core extension in the climb plane.

49x35mm (600 x 600 DPI)



Diffusion of a vacancy in a 2D-quasiperiodic structure: (a) the vacancy (open square) occupies a node site of the quasiperiodic lattice. (b) The vacancy and one atom exchange their sites. (c) The vacancy moves away to another position of the quasi-lattice, but the atom sits at an intermediate position (open circle). This site does not correspond to an actual node position of the perfect quasiperiodic lattice and represents a so-called phasonic defect, referred to as 'basculon'. In that case, vacancy diffusion enhances phasonic defect nucleation, allowing a reduction in the atomic flux associated with vacancy migration.
58x145mm (200 x 200 DPI)